

An Evaluation of Power Law Breakdown in Metals, Alloys, Dispersion Hardened Materials and Compounds

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AN EVALUATION OF POWER LAW BREAKDOWN IN METALS, ALLOYS, DISPERSION HARDENED MATERIALS AND COMPOUNDS

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Abstract

Creep at high stresses often produces strain rates that exceed those that would be predicted by a power law relationship. In this paper, we examine available high stress creep data for pure metals, solid solution alloys, dispersion strengthened powder metallurgy materials and compounds for power law breakdown (PLB). The results show that, if PLB is observed, then the onset of PLB is generally observed at about $\dot{\epsilon}/D_{\text{eff}} = 10^{13} \text{ m}^{-2}$, where D_{eff} is the effective diffusion coefficient incorporating lattice and dislocation pipe diffusion. The common origins of PLB for the various systems studied can be found in the production of excess vacancies by plastic deformation. Anomalous behavior in two pure metals (nickel and tungsten) and a solid solution alloy (Fe-25Cr and Fe-26Cr-1Mo) has been analyzed and provides insight into this excess vacancy mechanism. In metal systems, the onset of PLB is related to a change in the nature of the subgrain structure developed. In the PLB region, subgrains become imperfect containing dislocation tangles adjacent to the sub-boundary, and dislocation cells are evident. The dislocation tangles and cells are the source of excess vacancies and increase the creep rate above that predicted from power law creep. If subgrains do not form then PLB is not observed. In solid solution alloys, in which the dominant deformation resistance results from the interaction of solute atoms with moving dislocations, excess vacancies influence the diffusion of these solute atoms. PLB is not observed in many systems. This is attributed either to the presence of a high equilibrium vacancy concentration (because of a low activation energy for vacancy formation) or to the inability to form subgrains.

Introduction

Creep at high stresses often produces strain rates that do not follow a power law relationship between strain rate and stress. At low stress, a power law relationship is observed, while, at high stress, greater strain rates are observed than would be predicted by this relationship. This phenomenon is referred to as power law breakdown (PLB). A good example of PLB is available from the work of Luthy, Miller and Sherby [1] on pure polycrystalline aluminum. The data from this study are shown in Fig.1 and plotted as $\dot{\epsilon}/D_{\text{eff}}$ versus σ/E , where $\dot{\epsilon}$, D_{eff} , σ , and E are strain rate, effective diffusion coefficient, stress, and Young's modulus, respectively. The data show a PLB region which extends for 15 orders of magnitude in $\dot{\epsilon}/D_{\text{eff}}$ above the point that can be represented by a power law function. Garofalo [2] has shown empirically that for a number of metals creep data can be represented by a hyperbolic sine relationship between strain rate ($\dot{\epsilon}$) and stress (σ). The equation proposed by Garofalo is

$$\dot{\epsilon} = K (\sinh A \sigma)^n \quad (1)$$

where K , A and n are constants, with K a function of temperature. The hyperbolic sine relationship predicts a power law dependence between strain rate at low stress but transitions to an exponential relationship at high stress. A good example of the application of Equation (1) is the aluminum data shown in Fig. 1. The Garofalo equation has been fitted to this data (represented as the solid line in Fig. 1) and shown to provide excellent agreement between experimental data and the equation over 21 orders of magnitude in $\dot{\epsilon}/D_{\text{eff}}$. Power law breakdown appears at $\dot{\epsilon}/D_{\text{eff}} = 10^{13} \text{ m}^{-2}$.

Much of our knowledge of PLB, including mechanistic understanding and quantitative treatments of high stress creep data where PLB is observed, was obtained in the 1960s. Since that time significant additional data has been obtained in the high stress region where PLB can be expected. The purpose of this study is to examine these data (primarily for metals and solid solution alloys) relative to the available theories. The results provide insight into the capabilities and limitations of existing theory and provide insight into the origins of PLB in various material systems.

Mechanistic Considerations

Mechanistic studies of creep at high stresses have, in general, attributed PLB to one of two different origins: (1) inherent in the deformation mechanism or (2) resulting from a change in deformation mechanism upon increasing stress. Barrett and Nix [3] and Weertman [4] have developed theories in which PLB is inherent in the deformation mechanism and the relationship between strain rate and stress involves a hyperbolic sine. The Barrett and Nix theory developed rate equations for steady state creep based on the diffusion-controlled motion of jogged screw dislocations. The jogs were assumed to produce a dragging force on the dislocation which resulted in a rate equation containing a hyperbolic sine term involving the stress. The Weertman theory assumed climb-controlled dislocation creep. Steady-state creep is obtained due to the climb and annihilation of dislocations. The rate equation derived for this mechanism contains a hyperbolic sine term involving the stress raised to the 2.5 power.

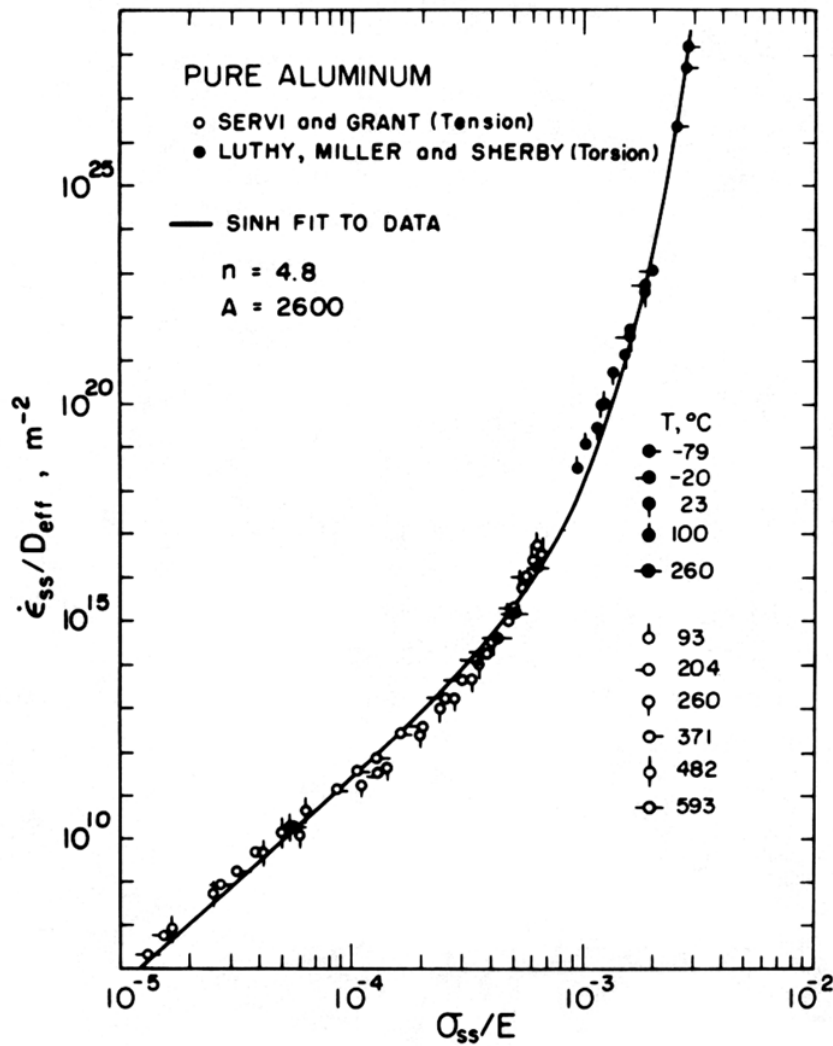


Figure 1. $\dot{\epsilon}/D_{eff}$ as a function of σ/E for aluminum (from reference 1). D_{eff} is the effective diffusion coefficient and E is the unrelaxed dynamic elastic (Young's) modulus.

Studies that have attributed PLB to a change in deformation mechanism have recognized that at lower stresses power law behavior is observed. The generally accepted mechanism is climb-controlled dislocation creep [4-6], in which dislocations climb to bypass obstacles [5] or are annihilated by dislocations of opposite sign [4]. At high stresses, where exponential behavior is observed, several mechanisms have been postulated including a transition from dislocation climb to glide [7], chemical interaction of jogs with supersaturation of vacancies, non-conservative motion of jogs, and jog migration and climb of edge dislocations. For alloys, in which the dominant deformation resistance is solute drag, Weertman [6] has proposed that, at high stresses, higher strain rates than predicted by a power law can be observed if dislocations break away from their solute atmospheres.

Power law breakdown can also occur through the production of excess vacancies, which, in turn, increases the diffusion coefficient. Excess vacancies are produced when the rate of vacancy production (due to sources such as dislocation climb or jog movement) exceeds the rate of vacancy annihilation (due to diffusion to sinks such as grain boundaries or subgrain boundaries). This mechanism applies to both climb-controlled dislocation creep

mechanisms described above (dislocation annihilation and obstacle bypassing) as well as the motion of jogged screw dislocations. In fact, both the Barrett and Nix and the Weertman theories consider the production of excess vacancies. For alloys, in which the dominant deformation resistance is solute drag, Weertman [6] has proposed that the production of excess vacancies can increase the diffusivity of the alloying elements, which can produce higher strain rates and thus PLB.

High Stress Creep Data

Creep data at high stress have been collected and studied for the presence or absence of power law breakdown. The materials analyzed include pure metals, solid solution alloys, dispersion strengthened powder metallurgy materials and compounds. For most materials, the creep data are analyzed as $\dot{\epsilon}/D_{\text{eff}}$, where $\dot{\epsilon}$ is the strain rate and D_{eff} is the effective diffusivity. For several of the materials (Ni, α -Ti, α -Zr, α -Ti, Al-3Mg, and DS (dispersion strengthened) Al-Mg), the data is represented as $\dot{\epsilon}/D_L$, where D_L is the lattice diffusivity. The quantity D_{eff} is defined as

$$D_{\text{eff}} = D_L \cdot f_L + D_P \cdot f_P$$

where D_L and D_P are the lattice diffusion coefficient and pipe diffusion coefficients respectively and f_L and f_P are the fractions of atoms participating in lattice and pipe diffusion respectively. Table 1 shows the materials studied, range of $\dot{\epsilon}/D_{\text{eff}}$ examined, stress exponent (n from $\dot{\epsilon} = k\sigma^n$), the value of $\dot{\epsilon}/D_{\text{eff}}$ at which PLB is observed and the references to the creep study. The results are plotted in Fig. 2. The range over which high stress creep data was obtained is indicated by solid line. If PLB is observed, this value of $\dot{\epsilon}/D_{\text{eff}}$ is indicated in the figure with a cross. The analyses presented here are based on the assumption that the diffusion-controlled creep process is the variable controlling PLB. It has been suggested by Raj et al. [37-39], however, that the value of the modulus-compensated stress controls the onset of PLB; at stresses above PLB the rate-controlling process in creep is assumed to be non-diffusion controlled.

The upper graph in Fig. 2 shows those materials that exhibit power-law breakdown. The materials are given in three subgroups: 1. High-stacking-fault-energy, subgrain-forming metals, 2. Solid solution alloys and 3. Subgrain-forming compounds. Most of the materials show approximately the same value of diffusion-compensated creep as observed with aluminum, $\dot{\epsilon}/D_{\text{eff}} = 10^{13} \text{ m}^{-2}$. The lower graph in Fig. 2 shows those materials that do not exhibit power law breakdown. These are also classified into three subgroups: 1. Anomalous diffusion coefficient metals, 2. Carbon steels and 3. Dispersion-hardened materials (constant structure creep). It is worth noting that powder metallurgy tungsten does not exhibit PLB (lower graph in Fig. 2) but cast ingot tungsten does (upper graph in Fig. 2). In the following section, we give an insight into possible explanations of the trends noted including the inconsistencies observed in nickel and tungsten.

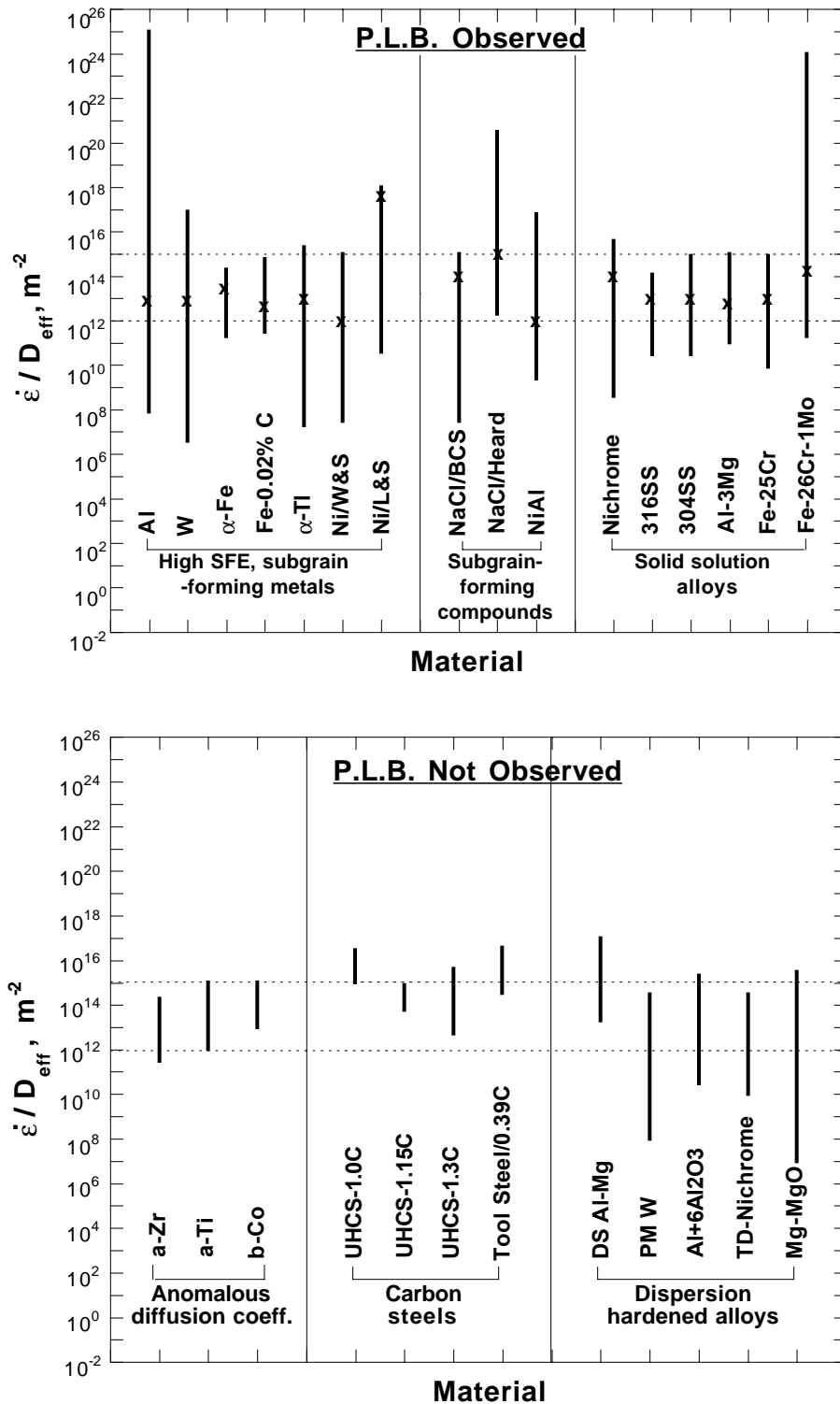


Figure 2. $\dot{\epsilon}/D_{eff}$ studied in high stress creep experiments for a number of materials. The results are divided into materials in which PLB is observed (upper plot) and materials in which PLB is not observed (lower plot). The range of $\dot{\epsilon}/D_{eff}$ studied is shown with a solid line. If PLB is observed, the value of $\dot{\epsilon}/D_{eff}$ at which PLB is observed, is indicated with a cross.

Table 1: High Temperature Creep Data

Material	Range of $\dot{\epsilon}/D$ (m ⁻²) studied	n	$\dot{\epsilon}/D$ (m ⁻²) at PLB	References
POWER-LAW BREAKDOWN OBSERVED				
High stacking fault energy, subgrain forming materials				
Al	8x10 ⁷ to 1x10 ²⁵	5 4.5	10 ¹³	[8] [1]
W	4x10 ⁶ to 8x10 ¹⁶	5	10 ¹³	[9]
α -Fe	2x10 ¹¹ to 1x10 ¹⁴	5	10 ¹³	[10]
α -Fe (0.02C)	3x10 ¹¹ to 6x10 ¹⁴	5	6x10 ¹²	[11]
α -Ti	2x10 ⁷ to 2x10 ¹⁵	5	10 ¹³	[12]
Ni	3x10 ⁷ to 1x10 ¹⁵	5	10 ¹³	[13]
	3x10 ¹⁰ to 1x10 ¹⁵	8	3x10 ¹⁷	[14]
Subgrain forming compounds				
NaCl	3x10 ⁷ to 1x10 ¹⁵	5	10 ¹⁴	[28, 29]
	2x10 ¹² to 3x10 ²⁰	5	10 ¹⁵	[30]
NiAl	2.5x10 ⁹ to 6x10 ¹⁶	5	10 ¹²	[31]
Solid Solutions				
Nichrome	3x10 ⁸ to 5x10 ¹⁵	5	10 ¹³	[15, 25]
316 SS	3x10 ¹⁰ to 1x10 ¹⁴	5	10 ¹³	[16, 31]
304 SS	3x10 ¹⁰ to 1x10 ¹⁵	5	2x10 ¹³	[17]
Al-3Mg	8x10 ¹⁰ to 1x10 ¹⁵	3.5	7x10 ¹²	[18]
Fe-25Cr	8x10 ⁹ to 8x10 ¹⁴	4	1x10 ¹³	[19]
Fe-26Cr-1Mo	2x10 ¹¹ to 1x10 ²⁴	4	3x10 ¹⁴	[19]
POWER-LAW BREAKDOWN NOT OBSERVED				
Anomalous diffusion coefficient				
α -Zr	3x10 ¹¹ to 2x10 ¹⁴	5	Not observed	[20]
α -Ti	1x10 ¹² to 1x10 ¹⁵	5	Not observed	[21]
β -Co	1x10 ¹³ to 1x10 ¹⁵	5	Not observed	[21]
Carbon steels				
UHCS-1.0C	1x10 ¹⁵ to 3x10 ¹⁶	5	Not observed	[22]
UHCS-1.15C-1.2Si	6x10 ¹³ to 8x10 ¹⁴	5	Not observed	[23]
UHCS-1.3C	5x10 ¹² to 4x10 ¹⁵	5	Not observed	[24]
Tool Steel (0.39C)	3.5x10 ¹⁴ to 3.8x10 ¹⁶	3.5	Not observed	[32]
Dispersion Hardened Alloys				
DS Al-Mg	2x10 ¹³ to 1x10 ¹⁷	8	Not observed	[33, 34]
PM W	1x10 ⁸ to 3x10 ¹⁴	8	Not observed	[9]
Al+6 Al ₂ O ₃	3x10 ¹⁰ to 2x10 ¹⁵	8	Not observed	[25, 35]
T-D Nichrome	1x10 ¹⁰ to 3x10 ¹⁴	8	Not observed	[25]
Mg+MgO	1x10 ⁷ to 3x10 ¹⁵	8	Not observed	[25, 36]

Data Analysis

General observations on materials exhibiting PLB

The approximately common value of $\dot{\epsilon}/D_{\text{eff}}$ at PLB that is observed for the materials in Fig. 2 is significant and suggests a common mechanism to explain their origin. It is proposed that the most likely explanation is the generation of excess vacancies by plastic deformation. Evidence for excess vacancy generation during creep in the PLB region was found for tungsten by Armstrong et al. [37]. Recently, Murty et al. [27] have measured deformation-induced vacancy production in super-high purity aluminum using in-situ nuclear magnetic resonance and found increased concentrations above the equilibrium values in strain-strain rate-temperature regions where this effect is observed. No theory has yet been successfully developed to illustrate why the diffusion-compensated strain rate is an important quantity relative to excess vacancy generation. Vacancy generation by a chemical force on the dislocation gives too high a value of the modulus-compensated stress where PLB is observed [38]. The chemical force for high vacancy generation is reached only when σ is about $10^{-1} G$ (where G is the shear modulus $\sim 0.4E$), rather than the observed value of about $10^{-3} G$. Figure 2 shows two significant exceptions to the value of $\dot{\epsilon}/D_{\text{eff}}$ at PLB. These are for pure nickel and pure tungsten both of which have been extensively studied. Understanding these anomalies gives some significant insight into the factors influencing PLB which are now discussed.

PLB in pure metals

PLB in nickel: Power law breakdown is observed in nominally pure nickel. However, there are two conflicting values for the onset of PLB. The data of Weertman and Shahinian (W and S) [13] reveal the typical value of $\dot{\epsilon}/D$ at PLB observed in high stacking fault energy metals with $\dot{\epsilon}/D$ equal to 10^{13} m^{-2} . On the other hand, the data of Luton and Sellars [14] show PLB to occur four orders of magnitude higher at 10^{17} m^{-2} . The tests by Weertman and Shahinian were done in tension at low strain rates (10^{-8} to 10^{-3} s^{-1}) whereas the tests by Luton and Sellars were done in torsion at high strain rates (10^{-3} to 4 s^{-1}). Comparison of the two sets of data reveal that the creep rates, at constant stress and temperature, observed by Luton and Sellars were about an order of magnitude higher than those observed by Weertman and Shahinian. An explanation of the different results obtained in the two sets of experiments is given as follows. The work of W and S is in the range where diffusion-controlled dislocation creep dominates with the formation of subgrains as the microstructural feature that leads to the normally expected PLB value. The work of Luton and Sellars was done under conditions where dynamic recrystallization was observed (some subgrain formation, however, was also observed). Dynamic recrystallization, perhaps occurring readily by the presence of the non-uniform strain in torsion, would be expected to lead to the low strength observed in the Luton and Sellars material. Dynamic recrystallization involves the creation of new strain-free grains and their growth under stress. This process could lead to the overall enhancement of atom mobility in the lattice either by moving grain boundaries or by the generation of excess vacancies during the nucleation and growth of new grains. These latter processes can assist dislocation climb. The proposal of excess vacancy generation has support in that when PLB occurs in the torsion tested nickel material (at $\dot{\epsilon}/D = 10^{17} \text{ m}^{-2}$), the creep rate merges into the power-law breakdown creep region given by Weertman and Shahinian. At this value of diffusion-compensated creep rate, the creep rate is enhanced by a factor of 15 over the creep rate predicted by extrapolation of the Weertman-Shahinian power-law creep data. We attribute the extrapolated increase in creep rate to an increase in the vacancy concentration.

PLB in tungsten: The anomaly in tungsten is summarized in Figure 3. In this graph the effective-diffusion-coefficient-compensated creep rate is plotted as a function of the modulus-compensated stress. Two groups are shown. One group, given as curve A, involving three different investigations, obeys the diffusion-controlled dislocation creep equation with five power law behavior at low stresses and PLB at $\dot{\epsilon}/D = 10^{13} \text{ m}^{-2} (10^9 \text{ cm}^{-2})$. The other group shown in curve B, exhibits 7 power law behavior and does not show PLB behavior which is not typical of pure metals. The later behavior is typical of constant structure creep where subgrains are not developed, and where the creep strength is dependent on the barrier spacing involving either grain boundaries or particles as barriers. The difference between curves A and B is attributed to the purity of the material. Curve B was obtained for tungsten made by powder metallurgy processing whereas Curve A was obtained by ingot metallurgy processing (the exception was the data of King and Sell who stated that their PM tungsten was prepared by a proprietary process which resulted in "superior purity"). It was shown that the grain size was an important variable in the creep of powder metallurgy tungsten. The creep rate increased as the grain size was increased, with $\dot{\epsilon}$ proportional to L^2 where L is the grain size. The value of L varied by about a factor of five [9].

The anomalous behavior of nickel and tungsten described above leads us to consider an explanation of PLB which involves subgrain formation during creep. It is well accepted that subgrains form in the power law creep regime. Furthermore, it has been demonstrated that the subgrain size is a function of the stress, following a reciprocal relation of the form $\lambda = k/\sigma$ where k is a material constant and λ is the subgrain size. Subgrains are formed as a result of dislocation climb of deformation-created gliding dislocations. This rearrangement of dislocations at subgrain boundary sites apparently precludes the storing of excess vacancies that are created during dislocation glide. This results in the well-established observation that the rate-controlling process in power-law creep is controlled by lattice diffusion with no evidence for excess vacancy formation. That is, the activation energy for creep is equal to the activation energy for lattice diffusion. As one enters the PLB region, dislocations also form subgrains but it is proposed that they are less perfect having less time to rearrange, and do not have the ability to sweep out any excess vacancies that are created. That is, the subgrain boundary becomes an imperfect sink for excess vacancies. The result is that the rate of creep is higher than that predicted by the equilibrium lattice diffusion coefficient because excess vacancies are contributing to the rate-controlling step in the creep process. The present proposal assumes quite reasonably, that the rate-controlling process is dictated by dislocation climb in the dislocation structure existing in the vicinity of the subgrain boundary. The proposed model involving subgrains can explain the absence of PLB in the powder-metallurgy tungsten studies. In this case, dislocations generated during creep have grain boundaries as barriers which are considered as perfect sinks for dislocations and vacancies. No dislocation rearrangements are needed near the boundary. The rate-controlling process is apparently not assisted by vacancy-creating moving dislocations. As pointed out earlier, the creep rate is proportional to the grain size squared, an indication that the grain boundary is the principal barrier to plastic flow. Why subgrains do not form in the PM tungsten materials needs additional study. It may be related to the presence of impurity particles associated with powder metallurgy processing.

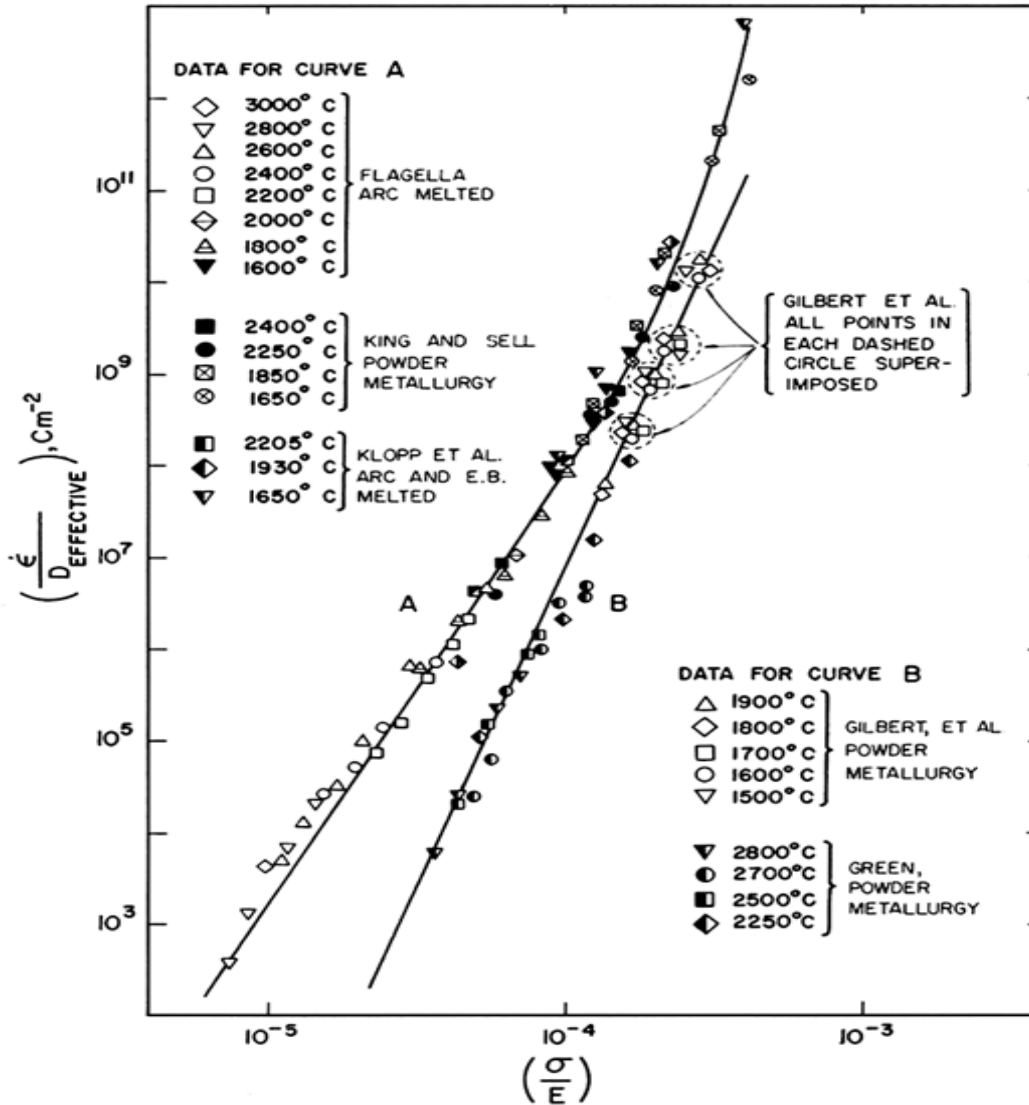


Figure 3. $\dot{\epsilon}/D_{\text{eff}}$ as a function of σ/E for tungsten (from reference 9).

PLB in compounds

PLB has been observed in sodium chloride and nickel aluminide. Both compounds show subgrain formation during creep. Two values of $\dot{\epsilon}/D_{\text{eff}}$ are listed for NaCl in Figure 2. The creep data of Heard [30] yield $\dot{\epsilon}/D_{\text{eff}} = 10^{15} \text{ m}^{-2}$ and those of Burke [28] yield $\dot{\epsilon}/D_{\text{eff}} = 10^{14} \text{ m}^{-2}$. These values are quite similar to those found for subgrain forming metals, and would suggest that a similar mechanism to that proposed above for tungsten and nickel is contributing to PLB in NaCl. The creep data reported by Raj et al. [39-41] on single crystal NaCl also indicate PLB at about the same value of $\dot{\epsilon}/D_{\text{eff}}$ as observed in the Heard and Burke studies. The dislocation structure developed in the single crystalline NaCl was analyzed in depth by Raj et al. [39-41] and the authors' conclusion was that a non-diffusion controlled process was operating at stresses above PLB related to glide polygonization, cross slip and recrystallization. The creep of NiAl has been analyzed extensively by Lesuer et al. [31] and by Raj and Farmer [42]. Both studies show power law and PLB behavior, with $\dot{\epsilon}/D_{\text{eff}} = 10^{12} \text{ m}^{-2}$. In agreement with the view of the present authors, Raj and

Farmer show subgrains as the dominant microstructure in the power law region and imperfect subgrains with dislocation tangles in the PLB region.

PLB in solid solution alloys

Power law breakdown is observed in solid solution alloy systems at essentially the same value of $\dot{\epsilon}/D_{\text{eff}}$ as for pure metals. It is concluded that a common mechanism can explain both cases. That is, excess vacancies are contributing to PLB. Solid solution alloys, however, have additional variables that may need to be accounted for in understanding PLB. The principal variable is that solute atoms can become the rate-controlling process by interacting with moving dislocations. The complex interaction of diffusing solute atoms with dislocations, the breakaway of dislocations from solute atmospheres (leading to inverse strain rate effects), the evaporation of solute atoms from dislocations at very high temperature, and the interaction of dislocations with stationary solute atoms are all factors that must be considered. Thus, it is possible that PBL in solid solution alloys occurs because of another rate-controlling process not directly related to lattice-diffusion-controlled dislocation creep. Figure 4 shows an example of the creep behavior for a ferritic stainless steel containing 25% Cr. The data are plotted as lattice-diffusion-compensated creep rate as a function of modulus-compensated stress. Two different stainless steels, based on Fe-25%Cr are shown, one containing 1%Mo and the other with no additional solutes added. Both stainless steels show power-law creep with a stress exponent of four; PLB occurs at about 10^{13} m^{-2} which is the same as that for pure metals. At higher values of $\dot{\epsilon}/D_L$, however, the creep rate is not only higher than predicted from power law extrapolation, but it reveals a negative diffusion-compensated strain rate effect. That is the stress exponent becomes negative. This is because of solute atom interactions with moving dislocations. It is readily apparent that solute contributions can sometimes interfere with using pure metal interpretations to explain PLB. Cottrell [26] has considered excess vacancy generation during creep as the basis for initiation of the Portevin-LeChatelier effect in Fe-C alloys occurring at $\dot{\epsilon} = 10^{13} \cdot D \text{ m}^{-2}$ where D is the diffusion coefficient of carbon in iron.

General Observations on materials that do not exhibit PLB

The lower graph in Figure 2 shows the materials that do not exhibit power law breakdown. An explanation of this observation is readily given based on the earlier discussion presented on materials showing PLB behavior. The first three materials shown in Figure 2 are α -Zr, α -Ti and β -Co. Two of the three metals show anomalous diffusion behavior. The activation energies for lattice self-diffusion in alpha zirconium and alpha titanium are one-half the predicted activation energies based on their respective melting points [43]. These low activation energies can be explained by a small activation energy for vacancy formation which would predict a high vacancy concentration. Thus, excess vacancy generation by deformation at high stresses will not occur until higher values of $\dot{\epsilon}/D_{\text{eff}}$ are achieved than those observed for normal metals. The case for β -Co cannot be explained in the same way since the lattice diffusion coefficient fits well with other fcc metals, although the reported data on dislocation pipe and grain boundary diffusion reveal wide range of values in the two reported investigations [44]. The highest value of $\dot{\epsilon}/D_{\text{eff}}$ plotted for β -Co of $9 \times 10^{14} \text{ m}^{-2}$ did show a slight upward deviation from the other data [21] so that PLB could have been occurring at about this value.

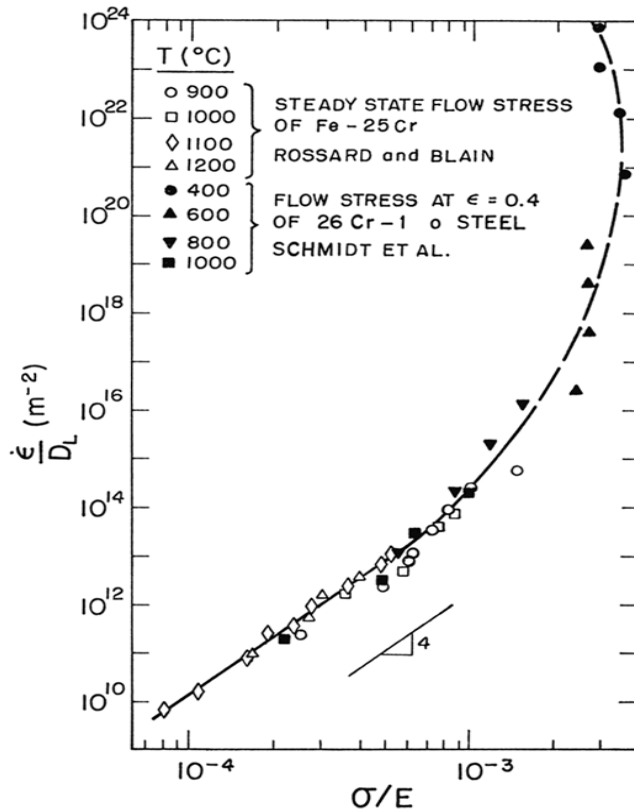


Figure 4. $\dot{\epsilon}/D_{eff}$ as a function of σ/E for 26Cr-1Mo steel and Fe-25 Cr (from reference 19).

No PLB was observed in ultrahigh carbon steels and in an Fe-C tool steel (Figure 2). The explanation could be related to the well-established influence of carbon on the self-diffusion of iron in austenite. Carbon in solution in austenite enhances the diffusion rate of iron (a factor of 20 from 0 to 1.4%C) and decreases the activation energy for diffusion. It is likely that the equilibrium vacancy concentration in Fe-C alloys is high, and this would require achieving PLB at higher stresses than those studied to date.

The dispersion-strengthened materials listed in Figure 2 do not exhibit PLB. This is in agreement with the present view that when subgrains do not form, PLB is not expected. Apparently the dislocation tangles and cells which are predicted as necessary to create excess vacancies are non-existent in most particle hardened materials. It is worth noting that particle-hardened materials do not exhibit primary creep and show steady state creep behavior from the beginning of plastic deformation. The barriers to plastic flow are not subgrains or cells but the closely spaced particles in the matrix. However, it is to be pointed out that solute interactions with moving dislocations in such materials are very important and represent the other major factor controlling the flow strength of such materials.

Concluding Remarks

The results in the previous sections show that, if PLB is observed, then the onset of PLB is generally observed at $\dot{\epsilon}/D_{eff} = 10^{13} \text{ m}^{-2}$. The common origins of PLB for the various systems studied can be found in the production of excess vacancies by plastic deformation. Anomalous behavior in two pure metals (nickel and tungsten) and a solid solution alloy (Fe-25Cr and Fe-26Cr-1Mo) has been studied and provides insight into this excess vacancy mechanism. In the pure metal systems, the production of excess vacancies, and thus the onset of PLB, was found to be influenced by dynamic recrystallization and sub-grain formation. The process by which dislocations form into

subgrains is recognized to be important. Imperfect subgrains can be imperfect sinks for vacancies, which will increase the vacancy concentration. Additional experimental and theoretical study of this concept is needed. In the solid solution alloys, in which the dominant deformation resistance results from the interaction of solute atoms with moving dislocations, excess vacancies influence the diffusion of these solute atoms.

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